Two-dimensional mapping of residual stresses in a thick dissimilar weld using contour method, deep hole drilling, and neutron diffraction

Wanchuck Woo[1], Gyu Baek An[2,*], Christopher E. Truman[3], Wenchun Jiang[4], and Michael R. Hill[5]

[1] *Neutron Science Division, Korea Atomic Energy Research Institute, Daejeon, 34057, South Korea*

[2] *Department of Naval Architecture & Ocean Engineering, Chosun university, Gwangju, 61452, South Korea*

[3] *Department of Mechanical Engineering, University of Bristol, Bristol, BS8 1TR, UK*

[4] *College of Chemical Engineering, China University of Petroleum, Qingdao, 266555, PR China*

[5] *Department of Mechanical and Aerospace Engineering, University of California, Davis, CA 95616, USA*

***Corresponding Author:**

Gyu-beak An

Department of Naval Architecture & Ocean Engineering,

Chosun University, Gwangju, 61452, South Korea

Phone: 82-62-230-7210

Fax: 82-10-6790-1204

E-mail: gyubaekan@chosun.ac.kr

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[*]Author to whom correspondence should be addressed; electronic mail:gyubaekan@chosun.ac.kr

Abstract

Residual stress variations were determined through the thickness of a 70-mm thick ferritic-austenitic dissimilar steel weld using contour method, deep hole drilling, and neutron diffraction. The result shows that significant tensile stresses were distributed distinctly along the interface between ferritic and austenitic phases. The band of the large tensile stresses was about 8 mm wide and the magnitude reached 400 MPa, which is approaching 100% of the yield strength of the base metal, near the top surface (about 15% of the depth). It is attributed to the large difference (5.8×10^{-6} 1/°C) of the thermal expansion coefficient between ferritic and austenitic steels of the interface. The microstructure analysis elucidates that the martensitic phase prevailed near the interface and results in microhardness increases.

Keywords: Residual stress, neutron diffraction, contour method, deep hole drilling, Dissimilar metal weld

1. Introduction

Most of the penetration nozzle components, steam generator systems, and large diameter pipelines in power plants and pressure vessels require dissimilar metal welds between ferritic and austenitic steels [1,2]. Although the dissimilar metal welds are widely used in many engineering structures, there are limitations and difficulties to predict the localized spatial variation of mechanical property and microstructures in such welds due to their inhomogeneity and the non-linear nature of the welding process [3-5]. In particular, it is critical to determine the location and magnitude of residual stresses in dissimilar metal welds because operational experience shows that serious cracking often initiates in the transition zone between the different materials, where residual stresses combine with applied loading and degrade material properties under extreme operating conditions [6,7]. A typical example of the dissimilar metal joining structure is the ferritic carbon steel (SA508) and austenitic stainless steel (316L) weld for the pressurized water reactor in nuclear power plants [8]. However, the high susceptibility to primary water stress corrosion cracking (PWSCC) in the dissimilar metal welds has been known to cause crack growth and may lead to abrupt fracture of key components [9].

A number of studies have been performed to determine the residuals stresses in the dissimilar metal welds based on computational simulation [10-16] and experimental methods [17-21]. Deng *et al*. predicted significant hoop stresses (over 140% of the yield strength of the weld metal) on the inside of a dissimilar metal welded pipe due in part to the large coefficient of thermal expansion (CTE) of the austenitic weld consumable (9.6Ni-19.9Cr-Fe bal.) [11]. Yaghi *et al*. reported that residual stresses were mostly significant near the top (outer diameter) surface in both the weld and the heat-affected zone (HAZ) of a 30 mm thick dissimilar metal weld joining two halves of ferritic steel pipes filled with an austenitic weld metal (1.48Fe-21.9Cr-Ni bal.) [14]. Eisazadeh *et al*. suggested that the primary role on residual stress formation is the CTE rather than the yield strength, thermal conductivity, and specific heat capacity in ferritic and austenitic (8Ni-18Cr-Fe bal.) dissimilar metal welds by systematic modeling studies [16].

Several experimental programs, complemented with simulations, have been focused on the residual stresses in welds at risk for PWSCC in nuclear power plant applications (ferritic to austenitic steel pipe joints with Ni-base alloy type weld metals) [17-21]. Joseph *et al*. measured residual stresses in 2.25Cr-1Mo ferritic steel and AISI 316 stainless steel pipes with and without Inconel 82 buttering using x-ray diffraction [17]. Kim *et al*. and Woo *et al*. used x-ray, hole drilling, and neutron diffraction to determine residual stresses in SA508 ferritic steel and 316L stainless steel pipes and weld overlay [18,19]. Ogawa *et al.* applied the deep hole drilling technique for the large scale (883 mm outer diameter) reactor vessel outlet nozzle to incorporate the residual stress distributions and the stress intensity factor in heavy-section structures [20]. Olson *et al.* recently reported extensive results from a full size nozzle (375 mm long and 35 mm thick) dissimilar metal weld by using the slitting, hole drilling, and neutron diffraction methods [21]. At this moment, it is important to measure the variations of residual stresses through the thickness of the dissimilar metal weld plate without any geometrical complexities. Furthermore, in order to elucidate the influence of the CTE difference on residual stresses, it is necessary to prepare a thick ferritic to austenitic dissimilar steel weld where the transient stresses can be fully developed along the interface during multi-pass welding.

In this paper, we present: (i) spatial variations of macroscopic residual stresses through the thickness of a 70 mm thick dissimilar metal weld specimen measured by three different methods (neutron diffraction, contour method, and deep hole drilling); (ii) comparison of the through-thickness stress distributions between the conventional similar metal weld (ferritic to ferritic) and the dissimilar metal weld (ferritic to austenitic) specimens; and (iii) the results of microstructure, hardness analysis, and residual stress dependency on CTE.

2. Processing, Microstructure, and Mechanical Properties

The base metal is the commercial high-strength low-carbon steel (wt% 0.05C, 0.1Si, 1.2Mn, 0.01P, and balance Fe). The average grain size was ~20 μm obtained by typical hot rolling at ~ 1150 °C, water quenching to 500 °C, and air cooling to room temperature. Two ferritic steel plates (each 600-mm long by 150-mm wide by 70-mm thick) were joined with an austenitic weld metal using multi-pass flux cored arc welding, Fig. 1(a). The austenitic weld metal was specially designed to not exhibit phase transformation, Table 1. The specimen was welded using a heat-input of 1.7 kJ/mm using a welding current, voltage, and electrode travel speed of 180 A, 29 V and 3.2 mm/s, respectively. The macroscopic structure is shown in Fig. 2 with cross-sections extracted from the plates. The welding process provided a bead width of about 60 mm on the top surface after 61 passes, with 21 layers welding in a groove of 30°, as shown in Fig. 2.

After welding, the dissimilar metal weld plate was cut slowly into two parts with the dimensions of 280 mm length (discard each edge of 20 mm) and 300 mm width by using a band saw, as shown in Fig. 1(a). One cut plate was provided for residual stress measurements using neutron diffraction, and the other for deep hole drilling and contour method. In the remainder of the paper x, y, and z directions denote longitudinal (welding), transverse, and normal directions, respectively, as shown in Fig. 1(a). Microstructural characterization was performed on the cross-section of the weld, Fig. 2. The locations for the optical microscopy were 5, 35, and 65 mm from the top surface along the weld centerline as marked by squares 1, 2, and 3 in Fig. 2. The microstructure of the weld metal exhibits a strong grain orientation, due to the elongated grains along the z direction (thickness) of the weld. The grain size is mostly over 200 μm at the face and center weld regions. Note that the strong texture and large grain sizes can cause significant errors or unavailability of data in diffraction phenomena of the neutron diffraction experiments.

Tensile specimens were machined from the base and weld metals at the mid-thickness with the gage length parallel to the longitudinal direction (x direction). Following ASTM E 8M-04, the tensile specimen was 6.25 mm diameter and 32 mm long in the gage section. The specimens were prepared using electrical-discharge machining (EDM) and tensile tests were performed at room temperature using a constant crosshead velocity providing an initial strain rate of 6.7 x 10^{-4} s^{-1}. The yield and tensile strengths of the base metal were 410 and 520 MPa and those from the weld metal were 460 and 630 MPa, respectively, as shown in Fig. 3(a) and Table 2.

Thermal dilation experiments were performed using samples of 3 mm diameter and 10 mm length cut by EDM from the base and weld metals. Thermal expansion and contraction were recorded during heating to 1300 °C at a rate of 1 °C/s, holding for 5 min, and cooling down to room temperature at a rate of 1 °C/s. The dilatometer test provides the CTE as 12.6 (1/°C, x10^{-6}) for the base metal and 18.4 (1/°C, x10^{-6}) for the austenitic weld metal by analysis of the linear expansion from room temperature to 100 °C. No phase transformation was observed in the austenitic weld metal during heating and cooling. Vickers microhardness (Hv) was measured on an etched weld cross section as shown in Fig. 2 at a set of locations from the weld centerline of 0, 30, 60, and 100 mm, all at 15 mm below the top surface. Fifteen microhardness measurements were collected near each location, and an average of the measurements was computed.

3. Residual stress measurements and data analysis

3.1. Neutron diffraction, contour method, and deep hole drilling

Three methods of residual stress measurement were used in this work: neutron diffraction (ND), the contour method (CM) and deep hole drilling (DHD). ND has become a well-established method for measuring macroscopic residual stresses in the interior of polycrystalline materials [22]. Spatially-resolved neutron strain scanning was performed by using the Residual Stress Instrument (RSI) at Korea Atomic Energy Research Institute (KAERI) [23]. The wavelength selection methodology, which minimized the total cross-section and the neutron beam attenuation, enables us to measure the residual stresses through the thickness of the 70-mm thick weld, Fig. 4 [24]. Wavelengths of 2.39 Å were selected for the diffraction planes (110) for the bcc ferritic base metal and (111) for the fcc austenitic weld metal at scattering angles of 71.4° and 72.0°, respectively. Although the wavelengths of (211) for bcc and (311) for fcc are generally recommended, the current diffraction planes were selected to maximize the available penetration length for the bcc ferritic and the fcc austenitic 70 mm thick weld metal, respectively [23]. Nominal scattering volumes of 4(x) × 8(y) × 4(z) mm^3 were used for diffraction with scattering vectors along the x direction, and a volume of 20(x) × 4(y) × 4(z) mm^3 for scattering along the y or z direction. Note that relatively large gauge volume is necessary to satisfy a criterion for the proper statistical peak profile analysis, i.e., the peak intensity (Ho) to background (Bo) ratio should be higher than 1.0 (>Ho/Bo) [22]. A total of 13 points were measured through the thickness starting 5 mm from the top surfaces to 65 mm in 5 mm steps, and measurements were repeated at locations from the weld centerline of 0, 30, 60, and 100 mm, as shown in Fig. 1. Mostly, the measurement period was about 1 hour for each strain component achieving a strain uncertainty of about ±100 με.

Diffraction peaks were analyzed using a least squares Gaussian fitting method in the RSI data analysis program [23]. Once the peak position was determined, the elastic lattice strains (ε) were calculated using $\varepsilon=-\cot\theta(\theta-\theta_o)=(d-d_o)/d_o$, where the θ_o (d_o) and θ (d) are the diffraction angles (d-spacings) for the stress-free and stressed materials at each position, respectively [22]. Generalized Hooke's law was used to convert elastic strains (ε_x, ε_y, ε_z) to residual stresses (σ_x, σ_y, σ_z) along the three orthogonal directions (x, y, z). The diffraction elastic constants and Poisson's ratios were E_{111} of 247.9 GPa, v_{111} of 0.24 for the austenitic weld metal (0 mm location) and E_{110} of 225.5 GPa, v_{110} of 0.28 for the ferritic base metal (30, 60, and 100 mm locations) [22]. Comb-like "stress free" reference samples were extracted along each line of strain scanning as shown in Fig. 1(d). The combs were 10 mm long (x), 4 mm wide (y), and 5 mm deep (z), Fig. 1(d). The stress-free lattice spacing (d_o) was carefully measured with a gauge volume of 8 mm^3 (2 × 2 × 2 mm^3).

Secondly, deep hole drilling (DHD), a mechanical strain relief technique for measuring residual stresses [24], was performed at the weld centerline (0 mm), Fig. 1(a). The longitudinal (σ_x) and transverse (σ_y) stress components were calculated via the distortions of a reference hole created through the thickness of interest, Fig. 1(c). Note that the incremental DHD (iDHD), which utilizes repeated hole-diameter measurements in each incremental machining step, is applied for the high magnitude of the plastic relaxation from 20 to 50 mm depth during the standard DHD process.

Finally, the contour method (CM) was applied to determine the weld residual stresses over the weld cross-section, Fig. 1(b). The displacements occurred due to the relaxation of the internal stress are compared to an assumed flat surface contour and the longitudinal (σ_x) residual stresses are recreated using a finite element model [24]. The forces required to ensure the measured deformed surface is returned to its original position represent the residual stresses. The method provides a two-dimensional map having a regular resolution of 0.5 × 0.5 mm of the residual stresses normal to the cut-surface. The stress calculations used Young's modulus and Poisson's ratio were E = 219 GPa for the base metal and 184 GPa for the weld metal as summarized in Table 2.

4. Results

4.1. Residual stress measurements by using ND and DHD

The measured distributions of residual stresses are shown in Fig. 5 for the 70 mm thick ferritic-austenitic dissimilar steel welded specimens. Each figure shows the through-thickness variations of residual stresses through the four different measurement locations as shown in Fig. 1(a). The stress uncertainties were mostly less than ± 50 MPa. Overall stress profiles seem to be different between weld and base metals, Figs. 5(a) and (d), which were measured at 0 mm and 100 mm locations from the weld centerline, respectively. The weld (0 mm) and HAZ (30 mm), Figs. 5(a)-(b), show that the three stress components similarly fluctuate, with a sine-wave like distribution. Smith *et al.* reported a similar stress profile in the region adjacent to the heat-affected zone of 108 mm thick steel weld [25]. The variation of residual stresses at the weld center and HAZ (±200 MPa) is not significant relative to the yield strength of the weld metal (460 MPa). Meanwhile, the stress profiles at 60 and 100 mm from the weld center, Figs. 5(c)-(d), exhibit an "M" shape with the σ_x and σ_y in compression up to -400 MPa near the surfaces balanced with tension (~ 200 MPa) at depths of about 25 and 50 mm. These profiles are typical in hot-rolled and quenched thick steel plates [26].

Figure 5(a) shows residual stresses obtained from the DHD and iDHD measurements along the weld centerline of the dissimilar metal welded specimen, Fig. 1(c). It should be noted that the ND measurements are unavailable at a few locations (5, 10, 25, and 30 mm depths) along the centerline due to the insufficient peak statistic from the austenitic weld metal caused by the strong texture and large grain size as shown in the face and the center of Fig. 2. In both measurements, σ_x shows higher magnitudes (up to 270 MPa at 15 mm) than σ_y at most depths excepting the distinct compression near the weld root (~65 mm). The DHD results agree with the ND results in the weld metal region within ± 50 MPa difference, Fig. 5(a).

4.2. Two-dimensional distribution of residual stresses in the dissimilar thick metal weld

Figure 6(a) shows the two-dimensional map of σ_x measured on the cross-section of the 70 mm thick dissimilar metal weld with the contour method (CM). Uncertainty is about ± 30 MPa. Overall, the stress map shows high tension near the weld metal balanced by compression in the base metal. It should be mentioned that significant tensile stresses (up to 400 MPa, about 90% of yield strength) are distributed along the interface between austenitic weld and ferritic base metals. It is a distinct feature of the ferritic-austenitic dissimilar steel weld, Fig. 6(a), when compared to the conventional ferritic similar steel weld, Fig. 6(b), from reference 24. Note that the welding parameters including heat inputs, welding passes, and geometries are similar of the two welds. Detailed comparisons will be addressed in the discussion section. Compressive residual stresses (-160 MPa) exist near the weld root (55~70 mm), resulting in an angular distortion of about 1° downward, Fig. 6(a).

Figure 7 shows profiles of σ_x extracted from the CM mapping along the four through-thickness lines (at 0, 30, 60, and 100 mm) as marked in Fig. 6(a). The previous result of DHD was included with the gray line for comparison, Fig. 7(a). Overall trends for the CM profiles are similar to the ND results in the four locations, though CM can provide much higher spatial resolution (1 mm spacing) than ND (5 mm gauge volume) for data analysis. Despite some scatters of ND results, which can be attributed to uncertainty in ‘stress-free’ reference specimens associated with microstructure changes, the results from the three stress measurement techniques are in good agreement.

Figure 8(a) represents the specific locations of residual stresses above 328 MPa (80% of yield strength) and shows the maximum residual stress (400 MPa) developed along the interface from 10 to 15 mm from the top surface. Figure 8(b) shows the results of microhardness (Hv) taken at four locations from the weld centerline of 0, 30, 60, and 100 mm as shown in Fig, 8(a). Microhardness (210 H_v) of the interface (HAZ 30mm) is higher than those (~170 H_v) of other locations.

5. Discussion

5.1. Residual stress comparison between similar and dissimilar metal welds

Let us discuss first about the difference in residual stress distributions between the dissimilar (ferritic-austenitic) and similar (ferritic-ferrritic) steel welds, Figs. 6(a) and 6(b), respectively. In both specimens, significant amounts of tensile residual stresses (over 90% yield strength of the base metal) were measured. In terms of the location, however, the significant tensile stresses were found near to the weld centerline at the top surface of the similar metal weld, Fig. 6(b), due to the accumulated thermal expansion/contraction and non-uniform plastic flow during welding [27]. Meanwhile, those were distributed near the interface between the weld and base metals in the dissimilar metal weld, Fig. 6(a). This is clear when comparing the stress versus profile position extracted from the maps 5 mm below the top surface, Fig. 6(c).

A number of computational simulations and experimental studies report that residual stresses are high near the interface of the dissimilar metal weld [10-21]. In general, it is considered that the relatively larger CTE with higher strain hardening rate (lower thermal conductivity and heat transfer rate) of the austenitic steel part induces higher tensile stresses after welding in dissimilar metal welds [15-17]. There is disagreement, however, regarding the location of the maximum stress, some reporting it in the austenitic steel part [10,16,17], some in the ferritic part [11,13,18,21], or on both sides [14,15]. The location of the tensile stress is important because it affects the crack initiation and fracture behavior of components

[28]. Although the yield strength of the ferritic base metal (410 MPa) is lower than that of the austenitic weld metal (460 MPa), Table 2, higher stresses were found in this work at the heat-affected zone, toward the ferritic steel region, as shown in Fig. 6(a). This highly stressed region is likely due to the significant strain hardening experienced via repeated welding processes followed by the fast cooling induced hard, brittle bainite and/or martensite structures in the multi-pass thick weld [29]. Indeed, the microstructures along the interface and transition zone, Fig. 2, exhibit localized banitic, tempered martensitic microstructures (arrows marked). Furthermore, Fig. 8(b) shows relatively higher microhardness of 210 H_v in the interface. The magnitude is consistent with the tempered martensite in 0.05 wt% C steels [30].

5.2. Residual stress dependency of the coefficient of thermal expansion.

It have been known that various parameters including yield strength, hardening modulus, thermal conductivity, specific heat capacity and/or geometries of components can affect residual stresses extensively in the dissimilar metal welds [11,14]. Systematic modeling studies by Eisazadeh *et al*. suggested that the CTE is dominant to determine the residual stress formation rather than the yield strength, thermal conductivity, and specific heat capacity in ferritic and austenitic (8Ni-18Cr-Fe bal.) dissimilar metal welds [16]. Deng *et al.* [11] and Lee *et al.* [13] emphasized that sufficient thermal stresses can be caused by the large CTE difference (ΔCTE) between low alloy steel and austenitic stainless steel. The thermal strain (ε^{th}) can be estimated based on a simple calculation (ε^{th} =ΔCTE×ΔT) [11]. The current ΔCTE of $5.8\text{x}10^{-6}$ 1/°C can cause a thermal strain of about 1700 με occurs as the weld cools from 300 °C to room temperature during welding. It is corresponding to 370 MPa with the elastic constant of 219 GPa in the ferritic steel and comparable to the residual stress, as shown in Fig. 6.

Figure 9 shows the relationship between residual stress and ΔCTE in ferritic-austenitic steel dissimilar metal welds. It was constructed from various prior results and the current experimental data. References 10 and 17 are from experiments and others are simulations. The normalized residual stresses were obtained by the maximum stress divided by the yield strength of the location. It shows a high correlation between the maximum residual stress and ΔCTE in the dissimilar metal welds, though there is a variance as the ΔCTE increases. Relatively higher residual stresses were found in the pipe weld cases [10,11,15] than plate welds [13,16]. It is likely due to the large constraint of through-wall bending moments in pipes [27]. Meanwhile, the residual stresses were reduced in refs. 14 and 17. Those cases use the welding consumable of Inconel type alloys, which have an intermediate CTE between austenitic and ferritic steels [14,17]. The current experimental data is located in the low side of the trend. It can be attributed to the relatively low heat-input (1.7 kJ/mm) and the tempering effect during the multi-pass welding of the 70 mm thick plate specimen.

6. Conclusions

1. Microstructure, longitudinal tensile properties, coefficient of thermal expansion (CTE), and residual stresses were extensively examined in a 70 mm thick dissimilar metal weld specimen joined with ferritic steel base metal and austenitic steel weld metal. The yield strengths of 410 and 460 MPa, and Young’s modulus of 219 and 184 GPa were obtained in the base and weld metals, respectively. The CTE difference was 5.8×10^{-6} 1/°C. No phase transformation was observed in the austenitic weld metal, whereas the martensitic phase was found near the interface between ferritic and austenitic steels and results in microhardness increases.

2. Nondestructive neutron diffraction measurements provide a total of 48 magnitudes and spatial distributions of the residual stresses along the three orthogonal directions of the

dissimilar metal weld plate. There were a few unavailable measurement locations in the austenitic weld metal due to the insufficient peak statistic caused by the strong texture (5-10 mm depth) and large grain size (25-30 mm depth). Destructive contour method and deep hole drilling measurements can complement the stress components by eliminating microstructure-induced complexities. The longitudinal stress results by the three methods are consistent within the difference range of ± 50 MPa.

3. Significant amounts of tensile residual stresses (approaching 100% yield strength of the base metal) were measured near the interface between the weld and based metals in the dissimilar metal weld. The residual stress mapping reveals that the maximum residual stress (400 MPa) was developed along the interface from 10 to 15 mm from the top surface (0.15~0.2 depth to thickness ratio).

4. The neutron diffraction (ND) provided the three stress components nondestructively at the specific locations of the specimens. The deep hole drilling (DHD) technique confirmed the longitudinal and transverse stress components along the hole penetrating along the weld centerline of the specimen. The contour method (CM) constructed the two-dimensional map of the longitudinal stress component perpendicular to the cut cross-section. There is good agreement in terms of the magnitudes and spatial distributions of the residual stresses among the three measurement methods.

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